Department of Metallurgy and Materials Science

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Cambridge, Massachusetts

RESEARCH ON MECHANISMS OF ALLOY STRENGTHENING

I. Alloy Strengthening by Fine Oxide Particle Dispersion
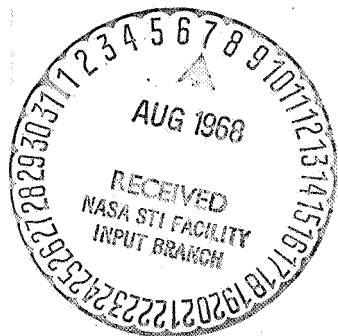
II. The Splat Cooling Process for Alloy Development

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INTRODUCTION

The two major aims of this program are the following:

(1) To study oxide dispersed alloys to establish the mechanism or mechanisms of strengthening, to examine in detail the modes of deformation and fracture under conditions of high temperature creep, to determine the nature of the structural stability which prevails and those factors which decrease the structural stability under conditions of long-time exposure at high temperatures, and to search for new and improved processing methods for the production of these alloys. It is felt that the powder metallurgy approach has not been examined to full advantage, and that a number of the chemical deposition processes are too involved and too expensive for general commercialization. The latter criticism begins to become particularly obvious when one contemplates the production of highly alloyed systems.

(2) To determine the potential of rapidly quenched, atomized or splat cooled particles and flakes as a basis for the development of improved or radically modified alloy systems capable of achieving high strength values at both low and high temperatures, coupled with improved structural stability.

This report covers studies pursued during the first half of 1968. Two doctoral theses were essentially completed at the end of 1967 and the very early part of 1968; one of these studies is now being prepared in rough draft form for publication, and it is anticipated that the second doctoral thesis will be put into publishable form within the next several months.

Excellent progress has been observed in all but one of the programs, that program having been delayed as a result of the slow delivery of a number of special alloys necessary for the production of the oxide dispersed structures.

A summary of the results of the individual studies follows.

A. Oxide Dispersion Strengthened Copper-Aluminum-Al_2O_3 by the SAP Technique

The reader is reminded that finely atomized copper-aluminum alloys were attrited to produce fine flakes with an average thickness near 0.5 micron. The first series, the A alloys, were merely surface oxidized to produce aluminum oxide in proportion to the amount of aluminum in solid solution in the alloy. Any copper oxide which may have formed on the surface of these powders was subsequently reduced in hydrogen, after which the product was compacted, canned, sealed, and extruded at 740°C at an extrusion ratio between 12:1 and 22:1. Each of these alloys had aluminum remaining in solid solution due to incomplete oxidation, the amount of unoxidized aluminum increasing with the total content of aluminum in the alloy. This amount of aluminum varied between 0.65 in alloy A 1 to 6.86 in alloy A 3. The presence of the aluminum in solid solution leads to significant

improvements in low temperature mechanical properties, and to excellent oxidation resistance at 450 and 650°C. Yield strength values as high as 83,000 psi and ultimate tensile strength values as high as 92,000 psi were reported, with 7.5 percent elongation and 10.5 percent reduction of area.

In creep at 450 and 650°C, these materials were not as good as alloy B 33 reported a number of years ago by Preston and Grant, a pure copper matrix containing 3.5 volume percent aluminum oxide, but the values were better than those reported for any other oxide dispersion strengthened alloy, and considerably better than the properties reported for any conventional copper alloys at these same temperatures. The reader should refer to the previous semi-annual report for specific values of strength at both low and high temperatures.

The B series of alloys in the program were prepared by attrition in the same way as were the A alloys. For the B alloys, the copper oxide on the surface of the fine, sub-micron flakes was utilized for further oxidation of the aluminum by an internal oxidation process. The oxidized fine flakes were exposed to an argon atmosphere for one hour at 725°C, permitting the copper oxide to dissolve into the alloy, thereby increasing the content of aluminum oxide and decreasing the content of aluminum remaining in solid solution. Not only does this procedure increase the amount of aluminum oxide, but it has the obvious advantage of decreasing the interparticle spacing by avoiding unoxidized regions in the centers of the fine flakes.

The compositions of the B series of alloys as well as those for alloys A 4 and B 4, a material containing 15 percent nickel as a solid solution strengthening element, are shown in Table I. It will also be observed that the extrusion ratio for these alloys is 22:1, except in the case of B 3 which had an extrusion ratio of 30:1. The extrusions were all sound and satisfactory. Density measurements indicate that each of the alloys shows better than 99.6 percent of theoretical density.

TABLE I

Composition of Alloys after Surface Plus Internal Oxidation

Alloy No.	Weight Percent				Extrusion Ratio
	Al	Al_2O_3	Ni	(v/o Al_2O_3)	
A 4	.33	.72	15.0	2.6	22:1
B 1	.51	.88	–	3.1	22:1
B 2	2.65	1.27	–	4.2	22:1
B 3	6.54	1.68	–	6.0	30:1
B 4	0.09	1.19	15.0	4.0	22:1

Metallographic examination of the alloys showed the oxide to be well dispersed throughout the material. Hardness versus annealing temperature tests were performed on all alloys; the results are plotted in Figure 1. Reference to earlier hardness plots allows comparison between these materials and the ones produced by surface oxidation alone. Each of the B alloys show a hardness drop at temperatures lower than for the comparable A alloys. The explanation for this behavior is not yet clear; metallographic and other studies are planned to determine the cause of the hardness drop at relatively low temperatures.

Tensile testing at room temperature and 450°C (842°F) has been completed; the results are shown in Table II. All testing was done on an Instron tensile testing machine at a constant crosshead speed of .05"/min. The A series alloys are included for completeness.

TABLE II

Tensile Tests at 20° and 450°C for A and B Type Alloys

Temp., °C	Alloy	0.2% Yield Stress, psi	UTS psi	Elong. %	Reduction of Area %
20	A 1	46, 200	49,700	19.8	22.6
	A 2	53,400	59,600	19.6	26.4
	A 3	83,100	92,600	7.5	10.5
	A 4	75,300	75,300	5.1	8.0
	B 1	68,300	70,000	12.1	17.8
	B 2	66,000	68,600	4.8	6.1
	B 3	83,900	88,700	4.6	2.1
	B 4	40,400	44,400	17.4	29.5
450	A 1	23,500	23,800	7.7	8.6
	A 2	28,200	29,600	9.2	8.9
	A 3	32,800	34,000	15.9	15.4
	A 4	25,400	25,400	6.1	4.8
	B 1	35,500	35,500	4.8	7.2
	B 2	38,500	38,500	4.2	5.0
	B 3	47,000	47,000	14.2	13.3
	B 4	23,800	24,200	12.0	15.2

The improvement in the 450°C tensile strength for the B series alloys (except for B 4) compared to that for A series is significant and will hopefully persist in stress-rupture properties.

Stress-rupture testing is in progress. The data on the B series as of now are too sparse to be included in this report.

One of the A series alloys, A 2, was cold swaged, 50 percent reduction in area, and subsequently machined to obtain stress-rupture specimens. The swaged material was stress-rupture tested at 450°C. The results of this testing are tabulated below and compared to results obtained for the as-extruded material.

TABLE III

Stress Rupture Test Comparison for Alloy A 2, As-extruded and 50% Cold Swaged

Material	Stress, psi	Rupture-time, hrs.
A 2 – As-extruded	25,000	1.35
	24,000	16.3
	23,000	25.8
	22,500	373.6
A 2 – 50% cold swaged	27,000	8.9
	25,500	103.0

These results give an estimated stress for 100 hours life of 23,000 psi for the as-extruded material and 25,500 psi for the swaged material. The benefits are significant. It should be emphasized that these results are preliminary and that a much more extensive program is under way to investigate the possible benefits of cold-work anneal cycles on these types of materials. The results of room-temperature tensile tests on the swaged material have been reported previously; they show an important increase in strength with cold work.

The following is an outline of work planned or already under way for the next period.

1) Stress-rupture testing of all alloys at 450°C and 650°C.

2) Swaging and annealing effects on room-temperature and high-temperature mechanical properties.

3) Determination of apparent activation energy for creep in various alloys and its dependence on composition and stress.

4) Production of another alloy of finer initial particle size than the ones already produced.

5) Correlation of accumulated data with data obtained from other studies to give a well rounded picture of the applicability of the SAP type method to the production of dispersion strengthened systems.

6) Structure examination by electron microscopy to determine the fine points of the resultant oxide dispersion.

B. The Role of Oxide Contaminants and Reactive Solute Metals in Oxide Dispersed Alloys

The element of greatest interest for the attainment of oxidation resistance is chromium. Unfortunately in all common oxidation resistant steels and superalloys about 18 to 20% Cr is required for protection up to 1800 - 2200°F. In producing fine metallic particles of near micron dimensions, considerable Cr_2O_3 forms which is reduced with much difficulty and is seldom lowered to less than 0.1% unless reduction temperatures of at least 2350°F are used (which leads to undesirably high degrees of sintering at fairly rapid rates.

A possible additional problem is a reaction between Cr in solution with the refractory oxide, leading to a small amount of reduction or interaction with the refractory oxide.

Emphasis has been placed on three heat resistant types of alloys, in part because of their potential for high temperature use, and in part by the availability of excellent fine powders of these alloys. The alloys are type 316 stainless steel, 80 Ni - 20 Cr alloy and 80 Fe - 20 Cr alloy. The refractory oxides have been ThO_2 and BeO.

Initial studies were made with as-received, 10 micron, type 316 stainless steel powders plus ThO_2. Hydrogen reduction of Cr_2O_3 at 2200°F lowered the content to about 1% Cr_2O_3, but led to extensive sintering which slowed down the reduction and finally stopped it off. Properties at 1800°F were poor; agglomeration of ThO_2 was observed.

As an alternate route for the production of heat resistant oxide dispersed alloys, the ONERA halide transport process was used to produce 80 Ni - 20 Cr - 3.5 % ThO_2, 45 Ni - 20 Cr - 30 Co - 5 Mo - 2.5% ThO_2, 80 Fe - 20 Cr - 3.5 % ThO_2; in addition Fe - 1.25 w/o BeO, produced by internal oxidation, was also chromized to introduce about 20% Cr. The resultant structures were relatively low in Cr_2O_3 but showed islands of a very rich chromium structure. The room temperature properties were excellent but the 1800°F values were extremely poor. In fact, an internally oxidized Fe-BeO alloy of the same BeO content, without chromium, was better in 1400°F tests. Surface chromizing of the Fe-BeO alloy in test bar form gave excellent stress rupture properties at 1400°F.

Subsequently two more type 316 stainless steel-thoria alloys (H3 and H5) were prepared. For these alloys the ThO_2 was added during attrition of the 316 stainless steel powder.

The resultant ThO_2 dispersion observed in the extruded bar stock was significantly improved; the $80°$ $\pm$ tension properties and the 1800°F stress rupture results also showed a small improvement.

To improve still further the interparticle spacing of the ThO_2, attrition time was increased by a factor of 4. The resultant powders were below 1 micron in thickness and on drying were observed to be pyrophoric; actual size determinations are under way.

Attrition is now being carried out in a stainless steel shell with 1/8" stainless steel balls. Recirculation of the powder during attrition has been initiated to avoid settlement of a coarser powder fraction in the bottom of the container.

Improvements have also been made to permit a check of the degree of completeness of the hydrogen reduction treatment. Based on these tests, instead of reduction at 2280°F for 4 hours, reduction is now carried out at 2200°F for 6 hours after which a check is made (Karl Fisher Reagent) of the exit gas according to a color change. If water vapor is indicated, reduction is continued; if the test is negative, the powder is ready for further processing. Six alloys have now been made, based on type 316 stainless steel, and tests are under way; the alloys are H-10 to H-15.

Table IV summarizes the alloys prepared to date. Table V lists the approximate values of stress for rupture in 100 hours at 1400, 1500, and 1800°F for earlier alloys. Table VI lists the content of ThO_2 and Cr_2O_3 for a number of the alloys. Table VII shows the 80°F tension data for most of the alloys studied thus far.

In the next six months, it is planned to put into effect the total improvements generated over the past 2 years to produce on the one hand attritor ground and blended type 316 stainless steel – ThO_2 material for extrusion and test; on the second hand improved halide decomposed alloys will be available to test as a comparison with the product made by attrition of powders.

TABLE IV

Summary of Oxide Dispersed Heat Resistant Alloys

Alloy No.	Nominal Composition	Extrusion Ratio	Extrusion Temp.°F	Remarks
H1	Ni-20Cr-2.5 w/o ThO_2	22:1	1800	ONERA halide process
H2	45 Ni-20Cr-30Co-5Mo-2.5 w/o ThO_2	22:1	1800	ONERA halide process
H3	3.6 s.s.* - 6 w/o ThO_2	22:1	1800	Attrited for 1 day and extruded
H4	Fe-20Cr-1.25 w/o BeO	22:1	1800	Internally oxidized Fe-Be, chromized and extruded
H5	316 s.s. - 6 w/o ThO_2	22:1	1800	Attrited for 1 day, H_2 sintered and extruded
H6	316 s.s. - 6 w/o ThO_2	16:1	2000	As received powder (10 micron)
H7	316 s.s. - 6 w/o ThO_2	16:1	2000	As received powder (10 micron) H_2 sintered
H8	Ni-20 Cr-2.5 w/o ThO_2	22:1	2000	ONERA halide process
H9	Fe-20Cr-2 w/o ThO_2	22:1	2000	ONERA halide process
H10	316 s.s.	30:1	1800	As attrited, 4 days
H11	316 s.s.	30:1	1800	As attrited, H_2 sintered at 2180°F, 6 hours
H12	316 s.s.	30:1	1800	As attrited, H_2 sintered at 2280°F, 6 hours
H13	316 s.s.- 6 w/o ThO_2	30:1	1800	As attrited
H14	316 s.s.- 6 w/o ThO_2	30:1	1800	As attrited, sintered for 6 hours at 1200°C
H15	316 s.s. - 6 w/o ThO_2	30:1	1800	As attrited, sintered for 6 hours at 1250°C

* s.s. = stainless steel

TABLE V

Stress for Rupture in 100 Hours for Early H-Alloys

Alloy	Test Temperature, °F	Stress for 100-hour life, psi
H1	1800	*
H2	1800	*
H3	1800	540
H4	1400	2600
H5	1800	940
H6	1800	430
H7	1800	960
H6, swaged 50%	1800	460
H7, swaged 50%	1800	1250
H3	1500	2900
H5	1500	2100
H6	1500	data not taken
H7	1500	1500
H7, swaged 50%	1500	3400

*   Severe scatter; values low.

TABLE VI

ThO$_2$ and Cr$_2$O$_3$ Contents of Selected Alloys

Alloy No.	Total O$_2$ wt. %	ThO$_2$ wt. %	Cr$_2$O$_3$ wt. %*	Remarks
H3	1.6	2.3	4.28	Attrited 316 s.s.
H5	0.89	2.3	1.98	Attrited 316 s.s.
H6	1.24	5.5	1.85	Attrited 316 s.s.
H7	0.98	5.5	1.03	Attrited 316 s.s.
H1	0.64	2.20	1.21	Halide decomposition
H2	0.68	2.63	1.17	Halide decomposition
H4	0.78	1.23	nil	Chromized Fe-BeO

* Weight % Cr$_2$O$_3$ was calculated assuming that all oxygen in excess of that present as ThO$_2$ was combined as Cr$_2$O$_3$.

TABLE VII

Room Temperature Tension Test Data

Alloy	0.2% Y.S. X1000 psi	U.T.S. X1000 psi	Elongation %
H1	107.5	113.75	9.6
H2	162.5	167.5	4.7
H3	122.2	125.0	5.0
H4	72.5	104.0	16.6
H5	57.5	83.75	15.8
H6	92.0	123.5	17.2
H6, swaged 50%	87.0	128.0	20.7
H7	43.6	82.4	21.7
H7, swaged 50%	41.8	81.0	25.3
H8	113.75	126.25	6.7
H9	53.75	75.0	8.7

C. Internally Oxidized, Alloyed, Ni-BeO Materials

Four nickel base beryllium alloys in chip form were obtained from the Beryllium Corporation of America in May 1968. The nominal compositions are given in Table VIII.

TABLE VIII

Compositions, Weight Percent, of Ni-Be Alloys

Alloy No.	Be	Mo	Co	Ni
1	0.15	–	–	bal.
2	0.32	–	–	bal.
3	0.32	4.6	–	bal.
4	0.34	4.46	29.41	65.79

The chips, unfortunately very coarse, were broken down by hammering and then wet ball-milled in alcohol. Two ball mills of (7" dia. x 10 1/2") of stainless steel have been fabricated for this purpose; one will utilize stainless steel rods of three sizes (1" D x 1", 3/4" D x 3", 1/2" D x 1/2"), and the other will utilize hardened ferritic stainless steel balls of 3/4" and 1/2" diameter. The rods for the first mill have been cut, the corners rounded off, and polished in fine silicon carbide paste, and later with sawdust. The stainless steel balls have been delivered.

About three pounds of -100 mesh powder have been made from chips of alloy 1, in about 3 weeks, by the rod mill. The ball mill is expected to give better yield.

The ball milled powder will be attrited to about one micron size in a one-gallon laboratory model attritor No. 1-S. Some modifications to the existing unit have been made to get better size distribution and efficient operation. A liner made of Inconel 600, of one gallon capacity, has been fabricated to minimize contamination of the powder. A Jabsco pump (model 2620; 3/8") has been installed to circulate the slurry of powder and alcohol to get better and more uniform grinding action.

After attrition the powder will be internally oxidized at several oxidation temperatures; for this purpose a new apparatus has been designed and is being assembled. This apparatus will utilize a fluidized bed technique for oxidation. A Vycor tube will be

heated in a split furnace to the required oxidation temperature; the powder will be suspended by an upstream of gas coming from the bottom of the tube. Internal oxidation will be obtained by adjusting the ratio of hydrogen and water vapor, which in turn will control the partial pressure of oxygen. Argon will be used as a carrier gas. The flow meters for the set-up have been made and calibrated. The furnace has been installed and temperature controller has been checked. The Vycor tube has been specially made. The system is about 80% complete. All the parts are in house, and could be in operation this month. For each planned extrusion, about two pounds of attrited powder will be required. The internal oxidation apparatus is expected to be in operation by the end of July.

II. The Splat Cooling Process for Alloy Development

A. Splat Cooling on a Continuous Basis

The following improvements have been made in the splat equipment so that now a successful run can be made practically every time:

1. A resistance wound furnace was built to replace the gas-air furnace. This permitted over a pound of alloy to be melted for each heat with better control of the temperature of that melt. Also, alloys with higher melting points than aluminum can be atomized.

2. The design of the atomization gun was improved to yield a finer and more consistent spray. It consists simply of a torch nozzle which is held in contact with a 1/4 inch stainless steel tube with an Al_2O_3 inner tube. The inner tube is replaceable, allowing different alloys to be splatted with the same gun. All but the top 1/2" of the gun is placed inside the furnace, eliminating most of the clogging problems due to freezing of the liquid metal at the atomization point.

3. A carbide blade was machined to replace the razor blades previously used. This carbide blade can be used over and over again without being dulled by the rotating wheel. The blade was also spring loaded to assure continual contact with the wheel.

4. As yet the steel wheel is not water cooled, but such a cooling system has been designed for use with copper and other higher melting alloys. With aluminum, the temperature of the wheel rises to about 50°C, which is insignificant compared to the splat temperature of 750°C. Cooling the wheel has had no observable effect on the microstructure of the splat material.

With the equipment in good working order about 3 pounds of alloy 2024 has been produced by rapid quenching. The powder has been sieved to obtain flakes of the same size and hopefully of similar structure and properties. The most abundant size, 2 to 150 microns, was then hydrostatically compacted and sent for extrusion at 150°C and an extrusion ratio of 20:1. The extrusions have not been made as of this date.

The microstructure of the splat material was observed by mounting powders in Clear Cast, polishing in the normal manner, and etching in Keller's reagent for 20 seconds. The microstructure consisted of aluminum dendrites and/or fine grains with any second phase entirely in the grain boundaries. The dendrite arm spacing was then measured from 500X micrographs of the 2024 material. Most particles had a dendrite size in the 1-5 micron range. This corresponds to a cooling rate of 10^5 to 10^3 °C/sec, respectively. The cooling rate was determined from a plot of log dendrite spacing vs. average cooling rate in the paper by Matyja, Giessen, and Grant. The cooling rate obtained compares favorably with the cooling rate of 3×10^7 °C/sec for aluminum alloys obtained in this laboratory with small charge (20 to 50 mg) splat machines, against a Cu substrate. The pure Al-Si alloy series has been delivered and will be prepared as splat cooled flakes. One Al-Si alloy was splat cooled and compared with the original air-atomized material. The air-atomized alloy had a dendrite size of 10-20 microns corresponding to a cooling rate of 10^2 to 10^1 °C/sec; the splat-cooled Al-Si alloy had a dendrite spacing of 1-5 microns or a cooling rate of 10^5 to 10^3 °C/sec. Thus the mass produced splat material has a grain size and cooling rate in between that of air-atomized powder and near micron thick splat-cooled material.

X-ray patterns were run on 2024 alloy filings and 2024 splat material. The samples were scanned from $2\theta = 20°$ to $2\theta = 60°$ to determine if any corrosion products were present in the splat material. (Splats made of other aluminum alloys were corroded by the SO_2 in the local atmosphere. Fortunately 2024 is a corrosion-resistant alloy and the X-ray patterns for the filings and splat material were identical.

The lattice parameter of the 2024 splat powder was measured and compared with the original 2024 alloy. The lattice parameter decreased from 4.053 Å to 4.046 Å due to splat cooling. This indicates that some copper has remained in solid solution as a result of splat cooling.

Splat cooling will now begin on the three Al-Si alloys of 3.9, 7.4, and 11.4 weight percent Si.

The splat material obtained will be sieved, canned, and extruded at about 150°C at 20:1, the exact conditions depending on the results of the 2024 extrusion. It is hoped that tensile and creep data will be available on both the 2024 and Al-Si alloys within the next six months. This will give a good idea of the effect of fine grain size, oxide dispersion, and supersaturated solid solutions on the strength and stability of aluminum alloys.

B. Splat Cooling of Al-Cu, Al-Fe, Al-Ni, and Al-Pd Alloys: Supersaturated Terminal Solid Solutions of Aluminum

The study of these potentially important solid solution alloys of aluminum has been continued and is now yielding excellent results after the introduction of a protective atmosphere during alloy melting and quenching. The elimination of atmospheric corrosion has resulted in a considerable solubility enhancement, probably by improved contact with the substrate surface.

Preliminary calculated values of the lattice parameters of metastable supersaturated Al-Pd and Al-Fe alloys are presented in Figure 2. All alloys were splat cooled to room temperature in an argon atmosphere against a copper substrate. The lattice parameters were calculated from X-ray diffraction measurements. Both the Al-Pd and Al-Fe lattice parameter vs. composition plots appear to be nearly straight lines with negative deviations from Vegard's law behavior. The Al-Pd curve indicates that the maximum metasolubility closely approaches the eutectic composition of 7.5 atomic percent Pd. The maximum meta-solubility of Fe in Al appears to be between 4 and 5 atomic percent, i.e., well beyond the 0.9 percent Fe eutectic. There is considerable variation of lattice parameter and hence in the degree of supersaturation observed in the Al - 5% Fe splats. This presumably reflects a very strong sensitivity to cooling rate and overall purity. The degree of supersaturation falls off rapidly above 5% Fe, closely paralleling the behavior previously seen in the Al-Si system.

Difficulty has been encountered in reproducing results in the systems: Al-Ni, Al-Co, and the higher copper content alloys in Al-Cu. Even in a protective atmosphere these alloys show some contamination from oxidation or corrosion. The appearance of a corrosion product previously seen in all splats made in air coincides with a diminished measured

metasolubility. The amount of this unidentified phase is significantly reduced by quenching in an inert atmosphere. Improvement of the vacuum system of the controlled atmosphere splat apparatus has further reduced the incidence of this phase. A consequent improvement in the reproducibility of measured metasolubilities is expected.

A summary of this work is to be presented at the AIME Fall meeting, 1968. The tremendously large supersaturation achieved (in the case of Al-Fe, by a factor X200) raises hopes of interesting dispersion strengthening possibilities, which will be investigated subsequently.

C. Study of Atmospheric Corrosion of Splat Cooled Al-Alloys

A study correlating the SO_2 content of the air during March, 1967, and the rate of corrosion of splat cooled foils has been completed; a paper based on this work has been submitted to AIME.

D. Survey of Supersaturated Solid Solutions and Metastable Alloy Phases

A survey of published and unpublished work on metastable terminal solid solutions and intermediate phases which are produced by quenching techniques from the melt has been completed. About 30 binary and over 30 ternary alloy systems have been treated; the report is to be published shortly and is available as a preprint now. Copies will be mailed to NASA.

E. Production of Rapidly Quenched Powders by Plasma Spraying Techniques

A Thermodyne plasma torch has been adapted for the spraying of fine powders onto a moving surface. Technical difficulties with the torch have not yet permitted usable powders to be obtained; excessive oxidation of Ni powders was encountered. Further tests are planned.
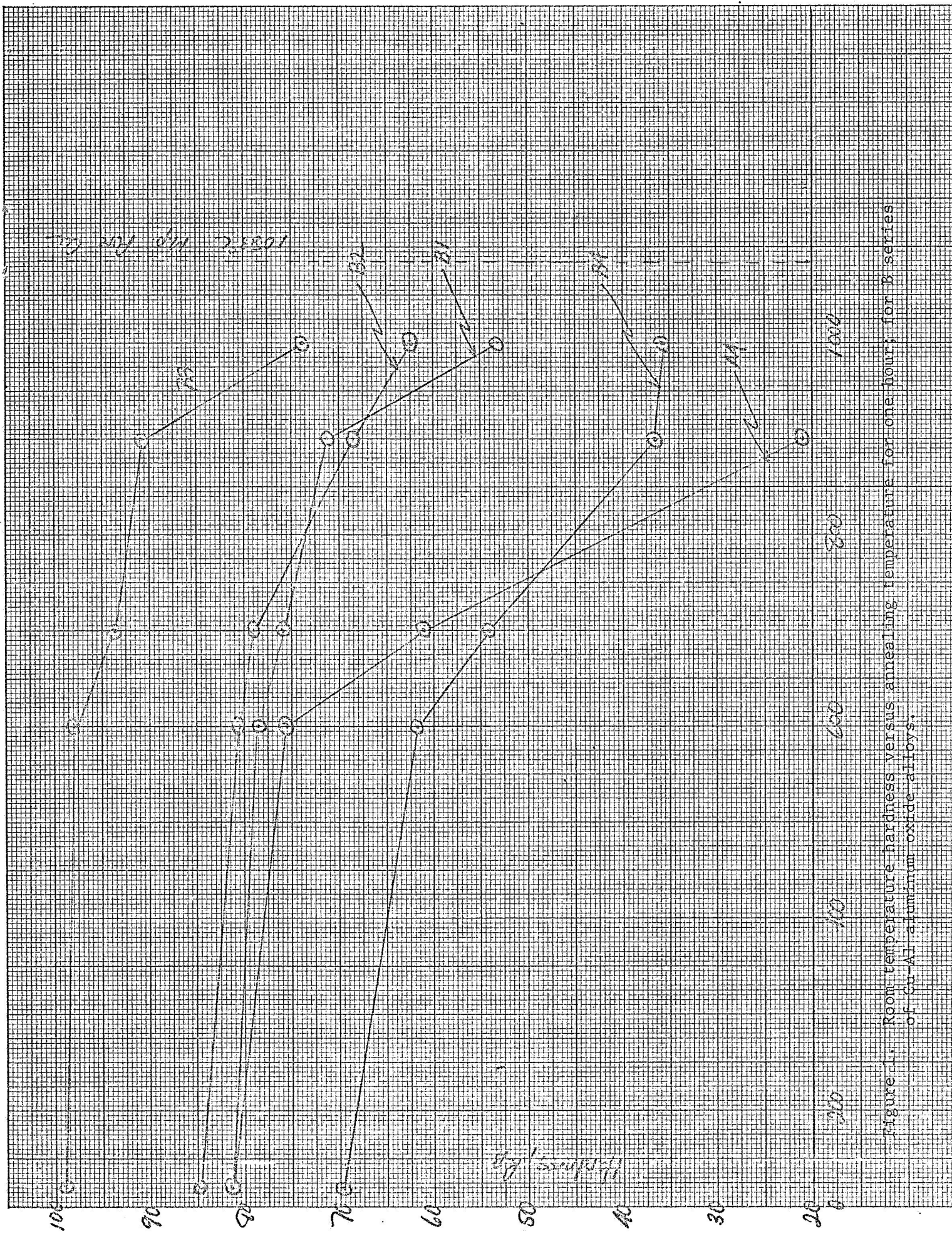
Figure 1. Room temperature hardness versus annealing temperature for one hour, for B series of Cu-Al aluminum oxide alloys.
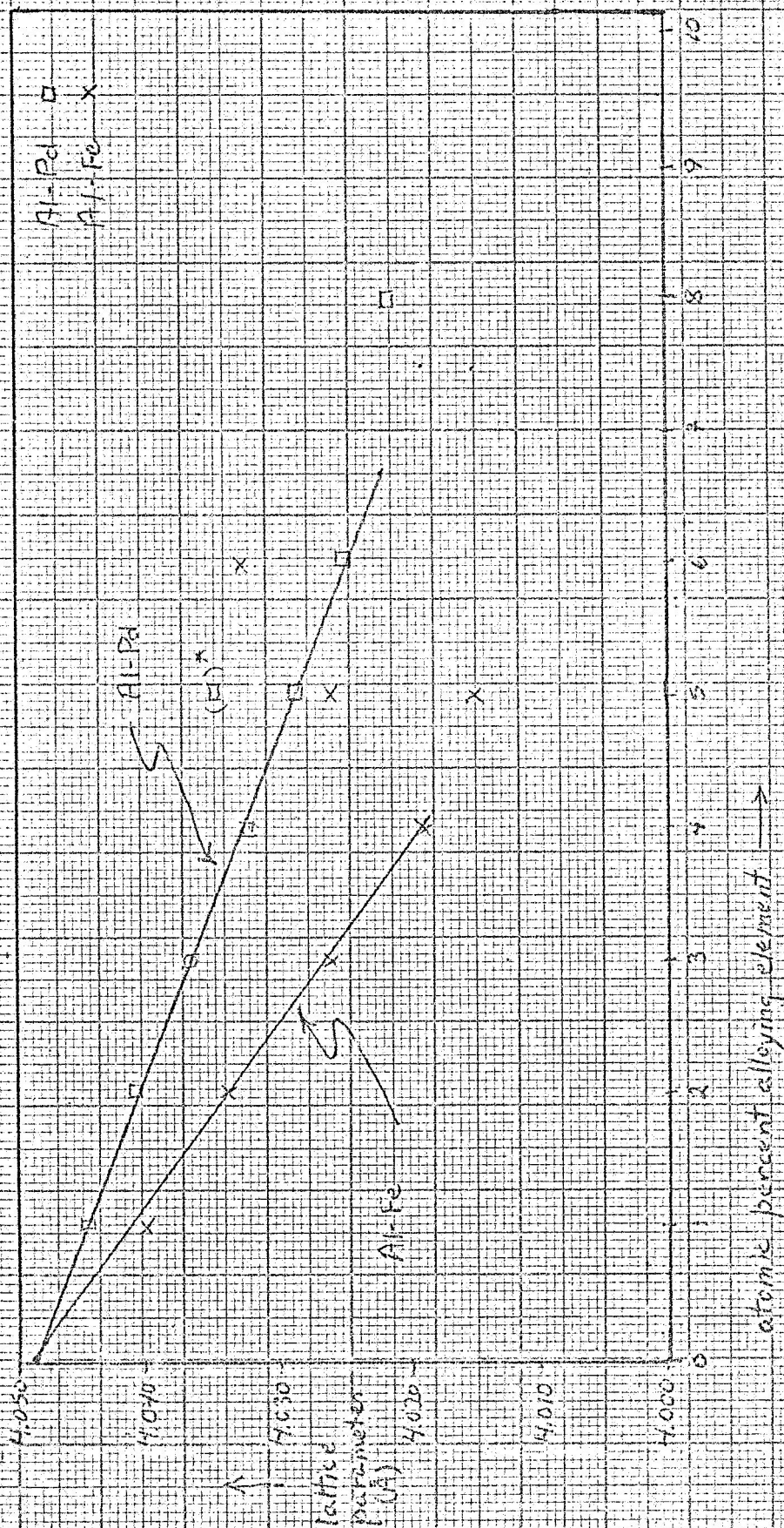
Figure 2. Lattice parameter vs. composition for the systems Al-Pd and Al-Fe. Alloys splat cooled to room temperature in an argon atmosphere. *Al - 5% Pd quenched to a temperature somewhat above room temperature.